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A Procedure for Investigating the Effect of Hydrogen Content on Toughness and Sustained Load Cracking Resistance of Titanium Alloys, With Some Results for Ti-6Al-4V

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13 ABSTRACT

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The present recommended hydrogen content limits of 125 ppm may prevent extreme embrittlement in the absence of cracks, but are of no value in reducing sensitivity to SLC. Contents below 35 ppm appear necessary to do that.

The substantial gain in toughness due to vacuum annealing a commercial heat of Ti-6Al-4V (R-23B), without reduction in hardness, raises the possibility of developing a vacuum annealing schedule for greatly increasing the reliability of critical high strength structural parts.

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INTRODUCTION

Titanium alloys are susceptible to sustained load cracking (SLC) due to residual hydrogen at hydrogen contents well below 100 ppm by weight. Experimentation at the Naval Research Laboratory has established that this cracking occurs at threshold stress intensity values, K_{IH}, well below the corresponding stress intensity for fast rising load fracture, K_{Ix} (or K_{Ic}, where this can be properly measured) (1-9). Because these SLC threshold measurements are similar in ambient air (70% relative humidity) (2) and high vacuum environments (7,9) for an alloy (Ti-8Al-1Mo-1V) which is extremely susceptible to stress corrosion cracking (SCC), it was concluded that SCC was not involved and that all titanium alloy SLC phenomena in air of ordinary humidity are caused by residual hydrogen.

From this background a useful method has been developed for evaluating the effect of hydrogen content on SLC properties in structural titanium alloys. The method utilizes the concepts of fracture mechanics (10), since effects on mechanical properties of hydrogen contents below 100 ppm are not evident unless precracked specimens are used. The hydrogen content of the specimens was controlled by vacuum annealing, followed by reannealing in hydrogen to achieve the desired hydrogen contents. Such stringent control is not required for practical evaluation of the effect of vacuum annealing on SLC for specific alloys or lots of one alloy, where reclamation of such material is all that is desired. Therefore, two procedures will be discussed, one dealing with reclamation of SLC-prone alloy lots, the other with the determination of specification limits for hydrogen contents in classes of alloys and types of applications.

PROCEDURE

Mechanics

For measuring the fracture toughness K_{Ix} or K_{Ic}, and threshold stress intensity, K_{IH}, for SLC of precracked specimens in air at room temperatures, the ASTM three-point bend specimen for fracture toughness measurement (11) is recommended as the simplest and most convenient to use. If it is desired to load the specimens as cantilever bars, the stress intensity calibration of Kies et al. (12) is appropriate in place of the ASTM three-point calibration. A bend specimen proportioned according to ASTM recommendations for 1-in. thick material is shown in Fig. 1. The specimens should be of sufficient size and of high enough yield strength so that the following criteria are met or exceeded (see Fig. 1):

$$B \geq 2.5 \left(\frac{K_I}{\sigma_{ys}}\right)^2$$

$$W-a \leq 2.5 \left(\frac{K_I}{\sigma_{ys}}\right)^2$$

$$a \leq 2.5 \left(\frac{K_I}{\sigma_{ys}}\right)^2$$

If these criteria are met, values of stress intensity will have been measured under plane strain conditions, and the recommendations for plane strain fracture toughness measurements (11) can be followed, so that K_{Ic} can be measured. K_{Ic} and K_{IH} can be used to calculate design criteria, such as critical crack sizes for fast fracture or sustained crack growth at design stresses. However, a valid determination of K_{Ic} is seldom possible for tough materials with moderate yield strengths. Hence, it is often necessary to be content with comparative measurements of K_{Ix} and K_{IH} for nonplane strain conditions, using standardized specimens with similar crack lengths. K_{Ix} is calculated using maximum load at fracture and the fatigue precrack size. Assuming B (Fig. 1) is fixed by plate thickness, W should be as great as possible but at least equal to 2B, S = 4W, and a = W/2.

Side grooves (shallow notches on each side parallel to the crack) are optional, but if used must be standardized throughout the test program. Side grooving will usually reduce K_{Ix} values, but may or may not affect K_{IH} measurements. Sandoz (1) found that for some samples of Ti-8Al-1Mo-1V 10-20% side grooving halved the thickness necessary to get minimum values of K_{IH} when there was a definite trend of decreasing K_{IH} with increasing thickness.

K_{IH} is estimated by first loading a specimen to about 1/2 K_{Ix}, waiting several hours, and then, if no crack growth indications are noted, increasing the load in 10 ksi$\sqrt{\text{in.}}$ increments, allowing each load to remain several hours. Some indicator of crack movement such as a deflection indicator or the ASTM crack opening displacement (clip) gage (11) should be monitored so that one can tell whether the crack is propagating, or whether the load should be increased at the end of some time interval. The lowest initial value of K_I resulting in sustained crack propagation and eventual fracture, based on the average crack length measured on the fracture surface (11), is recorded with the total time at

the load. This step loading procedure will provide accurate estimates of K_{IH} for specimens with ordinary commercial hydrogen contents, provided the specimen is never unloaded and reloaded. Two other specimens should be loaded to values of K_I which are 5 or 10 ksi$\sqrt{\text{in.}}$ below and above this value, and the times to failure (or to "run-out" if no failure occurs) recorded. In addition, times to failure for stress intensity values between K_{IH} and K_{Ix} are useful for estimating crack growth rates. This series of measurements is repeated for each hydrogen level to be investigated.

Thermometallurgy

A. Vacuum Dehydrogenation--The simplest procedure for determining the degree of vacuum annealing necessary to raise K_{IH} and K_{Ic} to some desired minimum value is to try a series of time-temperature schedules in the actual production vacuum furnace, or in an accurate laboratory prototype, and test as outlined above. If the specimens are to be heat treated after vacuum annealing, such heat treatment can be done in air, since the oxide layer which forms in air will prevent hydrogen from entering or leaving the specimen except possibly at very high temperatures and long exposure times. Generally, the highest temperature which will not metallurgically degrade the material, a good vacuum, and long times are needed to confer the greatest benefits. Ordinary vacuum facilities with ultimate pressures greater than 1 micron (10^{-3} mm Hg) are not suitable for the purpose, since the equilibrium hydrogen pressure exerted by a typical α-β titanium alloy containing 20 ppm hydrogen by weight, at 1600° F, is less than 1 micron (14). Consequently, the partial pressure of hydrogen from all sources in the system must be kept well below 1 micron to extract hydrogen from the alloy, because a hydrogen content of 20 ppm may well be too high (K_{IH} too low) for critical applications in some alloys. Since pressure readings are ordinarily made at some point remote from the specimen using gages calibrated for nitrogen, very low indicated pressures may be needed to attain acceptably low hydrogen contents after annealing. For example, specimens of Ti-6Al-4V held at 1700° F for 7 hours at an indicated total pressure of 2×10^{-6} mm Hg still retained 9 ppm of hydrogen. Tables in Reference 14 indicate that such a pressure at that temperature should have reduced the hydrogen content to less than 1 ppm within 2 hours. It has been found that the lowest hydrogen contents in that particular furnace are obtained by annealing at 1350° F for 40 hours, a time dictated more by the outgassing characteristics of the total system than by the time required for effusion of hydrogen from the alloy. A thorough study of Reference 14 is recommended before a vacuum degassing schedule is decided upon.

B. Method for Varying Hydrogen Content--The following procedure is suitable for varying the hydrogen content of titanium alloys in order to quantitatively evaluate the effect of hydrogen content on the fast rising load (K_{Ix}) and sustained load (K_{IH}) cracking resistance of titanium alloys. The initial step is to essentially dehydrogenate all specimens by vacuum annealing. K_{Ix} and K_{IH} measurements can then be made on the vacuum annealed specimens and on rehydrogenated specimens.

An apparatus for precision re-hydrogenation is shown in Fig. The vacuum annealed specimen is placed in the quartz tube, which is then evacuated and heated to 1700° F under vacuum. When 1700° F is reached, hydrogen is admitted, all valves are closed except No. 2, and the required pressure of hydrogen is bled into the calibrated volume via valve 3. The pressure is computed from the specimen mass (W, grams), the required hydrogen concentration (C, ppm), and the calibrated volume (V, cubic centimeters) from the following formula:

$$P \text{ (mm Hg)} = \frac{9.23\ WC}{V}$$

After the proper pressure is attained, valves 2 and 3 are closed, valve 1 is opened (4 remains closed), and the system (except the calibrated volume) is reevacuated. Valve 4 is then opened, all other valves are closed, and valve 2 is opened to admit hydrogen to the specimen. After the 0-760 mm gage reads essentially zero (a few minutes), valve 4 is closed. The system remains in this condition for 7 hours to ensure homogeneous distribution of the hydrogen in the specimen. The furnace is then allowed to cool to a temperature at which the specimen can be conveniently removed, or the quartz tube can be extracted from the furnace for faster cooling. 1700° F for 7 hours was selected as a condition which was fairly certain to give a complete absorption and uniform distribution of hydrogen in a 1-in. thick specimen. Experience indicates that a lower temperature, e.g., 1500° F, could be expected to give equally good results. The following points must be considered in selecting a treatment cycle: 1) Surface oxides must be completely dissolved, hence temperatures below 1300° F, at least for initial absorption, are risky. 2) The diffusion rate of hydrogen in titanium (which is different for α and β phases), is temperature dependent (14). For example, 7 hours at 1700° F is roughly equivalent to 30 hours at 1350° F, for absorption and homogenization. 3) The initial absorption time is not thickness dependent, but the achievement of a uniform hydrogen distribution depends on diffusion, and diffusion times are roughly proportional to the square of the thickness for a given temperature.

EXAMPLES OF RESULTS OBTAINED

Tests have been conducted on two heats of Ti-6Al-4V (Table 1), one containing 0.19% oxygen (R-23B), the other containing 0.07% oxygen (R-14), to show how hydrogen affects the sustained load (K_{IH}) and rapid rising load (K_{Ix}) crack propagation resistance of a typical high strength titanium alloy. The results are tabulated in Tables 2 and 3. Figure 3 summarizes the effect of hydrogen content on K_{Ix} and K_{IH}. All measurements were made on transverse (WR) specimens.

R-23B (0.19% oxygen) 145 ksi yield

Because the hardness of R-23B specimens was not reduced by any of the annealing treatments, and because the measured yield strength did not decrease after annealing in air at 1700° F for 7 hours and furnace cooling, it is believed that the yield strengths for all the conditions reported here were essentially the same.

The low toughness and SLC resistance of the as-received plate was substantially raised by simple annealing in air at 1700° F for 7 hours and furnace cooling. This toughness and SLC resistance were further increased by annealing the as-received material in a vacuum (5×10^{-6} mm Hg) at 1700° F for 7 hours and furnace cooling. Except for these effects, changes in hydrogen content over a wide range above 35-50 ppm had little effect on K_{IH} or K_{Ix}. The principal effect of increasing hydrogen contents above 34 ppm is to increase the sustained load crack growth rates, as indicated by the greatly reduced failure times near K_{IH} for specimens with higher hydrogen contents.

The reports in earlier work (13) that Ti-6Al-4V does not suffer embrittlement at hydrogen contents below 400 ppm were based on smooth or notched tensile specimens, which obviously are unsuitable for evaluating the effect of hydrogen content on toughness.

The few approximate values of K_{Ic} in Table 2 (specimens 289, 290, and 267) are for as-received and air annealed specimens. For these, K_{IH} is about 75% of K_{Ic}. This ratio is representative for as-received Ti-6Al-4V alloys which have been tested at NRL.

The specimens containing 122 ppm of hydrogen behaved erratically, possibly because they are sensitive to step loading (increasing the load after long intervals at previous loads). However, one specimen failed at a fairly low value

of K_I (No. 280) in spite of step loading, while the others did not fail until much higher K_I values had been imposed. There were no irregularities in the fatigue-precrack procedures or crack front shapes, nor was anything surprising discovered on the fracture surfaces, which were similar to those of all the other "K_{IH}" specimens. Although such behavior makes estimation of K_{IH} difficult, it is reasonable to place K_{IH} at 58 ksi$\sqrt{in.}$, where No. 280 failed, since this continues the trend established by the other hydrogen contents.

R-14 0.07% oxygen, 124 ksi yield strength as-received

The fast fracture toughness index, K_{Ix}, for R-14 decreases with increasing hydrogen content in the same way as for R-23B, but K_{IH} is not obviously affected. Although the specimens are under-sized with respect to the plane strain criteria mentioned earlier, it is hard to blame this factor for the lack of K_{IH} variation with hydrogen content, since K_{Ix} definitely increases with reductions in hydrogen in the same size specimens. Further reduction of hydrogen below 9 ppm might raise K_{IH}, because such a reduction (to 5 ppm) did have this effect on R-23B. That is, 9 ppm may be sufficient to fully "embrittle" R-14 with respect to sustained load cracking. The as-received toughness and SLC resistance of 1-in. thick specimens of WR orientation can only be inferred from measurements on 3/4-in. thick WT specimens by Beachem, et al. (15). Those results were consistent with those obtained on the air annealed RW specimens tested here. In contrast to its effect on R-23B, high hydrogen contents do not markedly increase crack growth rates in R-14. This, again, may be attributed to the lack of hydrostatic constraint (plane strain) afforded by the low yield strength at this thickness.

MISCELLANEOUS RESULTS

Annealing in air for 1 hour at 1700° F did not significantly raise the hydrogen content of vacuum annealed specimens of R-23B (Table 2, No. 294 and No. 295), and the hydrogen content of R-14 after annealing in air for 7 hours at 1700° F was 38 ppm, which is about equal to the producer's analysis (Table 1). On the other hand, the measured hydrogen content of R-23B specimens after annealing in air 7 hours at 1700° F was 65 ppm, compared with a measured 34 ppm in an as-received specimen. It is difficult to decide whether to blame variations in hydrogen content in the plate, or to conclude that prolonged annealing at high temperature in air increased the hydrogen content in R-23B, but not in R-14.

The chosen temperature, 1700° F, and annealing time, 7 hours, provided a completely uniform hydrogen distribution in all the specimens. Analyses for several locations from the surface to the center showed no consistent variations in hydrogen content within a given specimen. The absence of a hydrogen content gradient in vacuum annealed specimens signifies that the ultimate vacuum attained in the furnace dictated the hydrogen content of the specimens, and that hydrogen contents well below 8 ppm could be attained within 7 hours at 1700° F if the partial pressure of hydrogen at the specimens surface could be reduced.

DISCUSSION

The data presented here are based on 1-in. thick specimens, which are not large enough to satisfy the plane strain criteria presented in the beginning of this paper for R-14, but are satisfactory for all but vacuum annealed R-23B specimens. Although the numerical values of fracture toughness are not all useful for design purposes, the trends observed are useful and surprising, and point to some promising directions for titanium alloy metallurgical research. The "inevitable" embrittlement of Ti-6Al-4V by the addition of oxygen to raise the yield strength seems to somehow be tied to the presence of hydrogen. While K_{IH} of R-23B increased with a decrease in hydrogen content as expected, the toughness increased at least as much, without any observable decrease in hardness, and thus strength. The substantial increase in K_{Ix} and K_{IH} due to annealing at 1700° F in air, and the further increase in K_{Ix} and K_{IH} due to annealing in vacuum, without loss in hardness compared with the as-received plate, leads to two conclusions. First, commercial Ti-6Al-4V plate does not receive heat treatment calculated to optimize its toughness at high yield strength. Second, the combination of an optimum heat treatment with vacuum-dehydrogenation promises to yield a very high toughness while retaining high yield strengths.

The comparison between R-14 and R-23B is somewhat complicated by their differences in yield strength, but the superiority of the low oxygen content alloy is only apparent in these tests as a higher K_{IH} at moderate and high hydrogen contents. When both are vacuum annealed, their K_{IH} values are identical. The most important observation, however, is that both are much tougher after thorough dehydrogenation.

Although K_{IH} does not decrease with continued increase in hydrogen content above 35 ppm (in R-23B), the rate of cracking, reflected by time to failure above K_{IH}, does increase.

For example, the failure times of specimens with 206 ppm hydrogen change from hours to minutes over a span of 4 ksi$\sqrt{in.}$, near K_{IH}. SLC evidently is a stress dependent process, whose rate is controlled by the rate at which a critical local hydrogen concentration can be attained through diffusion to the region near the crack tip. It is possible that K_{IH} increases (in an operational sense) at very low hydrogen content because some competing process, such as creep-induced stress relaxation, forestalls cracking.

The present recommended limit for hydrogen in titanium alloys is 125 ppm. This limit is based on experience with plain or notched, but not cracked, tensile specimens, and evidently is of little value in controlling the sensitivity to SLC of titanium alloys. It serves mostly to prevent really appalling embrittlement problems from occurring. If SLC is to be eliminated, either hydrogen contents must be kept below 10 ppm, or some other approach to preventing SLC must be found.

The complex effect of hydrogen content on the toughness of R-23B cannot be explained with existing knowledge of titanium alloy metallurgy. In order to fully exploit the potential for good combinations of strength and toughness which may be attainable by as yet unthought-of changes in alloy chemistry, and thermomechanical processing, a more complete understanding of the possibly synergistic effects of interstitial and substitutional alloy elements and of processing techniques is necessary. In the context of the present results, the metallographic strengthening mechanism of oxygen, and the influence thereon of hydrogen, and possibly of other components, is a mystery which has never been truly explored, the understanding of which might make possible new trends in engineering alloy development.

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Table 1--Composition and Properties of Ti-6Al-4V Alloys

Alloy	Original Thickness (in.)	Chemical Composition (wt %)							Transverse Y.S. (ksi)		Hardness (R_c)
		Al	V	Fe	O	N	C	H	Vendor	NRL	
R-23B	1.02	6.5	4.2	0.18	0.18	0.012	0.02	0.0040	139	143	34.5
R-14	3.0	6.0	4.1	0.5	0.07	0.008	0.023	0.005	114	124*	32.0*

* Values from 1-in. thick plate specimens cut from exterior sections of 3-in. thick plate.

Table 2 - R-23B (0.18% oxygen) Experimental Results

I.D. No.	Thermal Treatment*	Hardness (R_c)	Hydrogen (ppm)	Est. Y.S. (ksi)	First Load (no cracking) K_I (ksi/in.)	First Load (no cracking) Time (hr)	Final Load (cracking) K_I (ksi/in.)	Final Load (cracking) Time (hr)	Comments
289	as-received	34.5	34	145	-	-	58	0	K_{Ix}·(K_{Ic} ~ 51)
290	as-received	-	34†	145	-	-	63	0	K_{Ix}·(K_{Ic} ~ 54)
292	as-received	-	34†	145	-	-	40	20.1	cantilever
303	as-received	-	34†	145	-	-	42	10.7	cantilever
266	1700Air7FC	-	65	145‡	-	-	86	0	K_{Ix}
267	1700Air7FC	34.0	61	145	-	-	87	0	K_{Ix}·(K_{Ic} ≈ 77)
273	1700Air7FC	-	65†	145	-	-	63	2.1	cantilever
281	1700Air7FC	-	65†	145	-	-	70	0.5	cantilever
282	1700Air7FC	-	65†	145	-	-	60	3.3	cantilever
297	1700Argon7FC	-	34†	145	-	-	72	0	K_{Ix}
305	1700Argon7FC	-	34†	145	-	-	103	0	K_{Ix}
258	1700Argon7FC	-	34†	145	-	-	60	8.9	3-point
296	1700Argon7FC	-	34†	145	-	-	70	0.0005	cantilever
306	1700Argon7FC	-	34†	145	51	25.9	61.5	19.1	cantilever
261	1700V7FC	34.0	8	145	-	-	132	0	K_{Ix}
265	1700V7FC	-	8†	145	-	-	132	0	K_{Ix}
270	1700V7FC	-	8†	145	-	-	129	0	K_{Ix}
257	1700V7FC	34.0	7	145	-	-	84	8.0	3-point
262	1700V7FC	-	8†	145	-	-	84	15.2	3-point
263	1700V7FC	-	8†	145	-	-	103	2.3	3-point
264	1700V7FC	-	8†	145	-	-	92	6.7	3-point
268	1700V7FC+1350V40FC	-	4.5†	145	-	-	129	0	K_{Ix}
269	1700V7FC+1350V40FC	34.0	4.5	145	72	6.4	99	56.4	3-point
295	1700V90FC+1700A1AC	-	11†	145	-	-	124	-	K_{Ix}
294	1700V90FC+1700A1AC	33.0	11	145	83	20.9	94	15.0	3-point
304	1700V7FC+1700$H_2$7FC	-	36†	145	-	-	105	0	K_{Ix}
291	1700V7FC+1700$H_2$7FC	-	36	145	-	-	110	-	K_{Ix}
293	1700V7FC+1700$H_2$7FC	34	36	145	-	-	59	19.8	cantilever
275	1700V7FC+1700$H_2$7FC	34.5	53	145	-	-	95	0	K_{Ix}
288	1700V7FC+1700$H_2$7FC	34	47	145	-	-	95	0	K_{Ix}
276	1700V7FC+1700$H_2$7FC	35	50†	145	-	-	61	6.1	3-point
302	1700V7FC+1700$H_2$7FC	-	50†	145	62	95.6	66	4.2	cantilever
279	1700V7FC+1700$H_2$7FC	-	122	145	-	-	91	0	K_{Ix}
310	1700V7FC+1700$H_2$7FC	-	122†	145	-	-	89	0	K_{Ix}
269	1700V7FC+1700$H_2$7FC	34	122†	145	47	1.25	58	0.2	3-point
286	1700V7FC+1700$H_2$7FC	35	122†	145	53	4.2	75	0.12	3-point
301	1700V7FC+1700$H_2$7FC	-	122†	145	60	21.5	81	0.033	cantilever
309	1700V7FC+1700$H_2$7FC	-	122†	145	63	2.1	70	5.3	cantilever
307	1700V7FC+1700$H_2$7FC	-	206†	145	-	-	91	0	K_{Ix}
318	1700V7FC+1700$H_2$7FC	-	206†	145	-	-	83	0	K_{Ix}
314	1700V7FC+1700$H_2$7FC	34	206†	145	-	-	64	0.17	cantilever
315	1700V7FC+1700$H_2$7FC	-	206†	145	-	-	64	0.13	cantilever
316	1700V7FC+1700$H_2$7FC	-	206†	145	-	-	62	2.7	cantilever
317	1700V7FC+1700$H_2$7FC	-	206†	145	-	-	60	4.3	cantilever

* 1700V7FC = Annealed at 1700° F in vacuum for 7 hours, then furnace cooled, for example.

† Estimated hydrogen content. No analysis available.

‡ Measured transverse 0.2% offset Y.S.

All K_{Ix} and K_{Ic} measurements 3-point bend. Maximum load reached in 10-15 seconds, crosshead speed 0.005 inches/second.

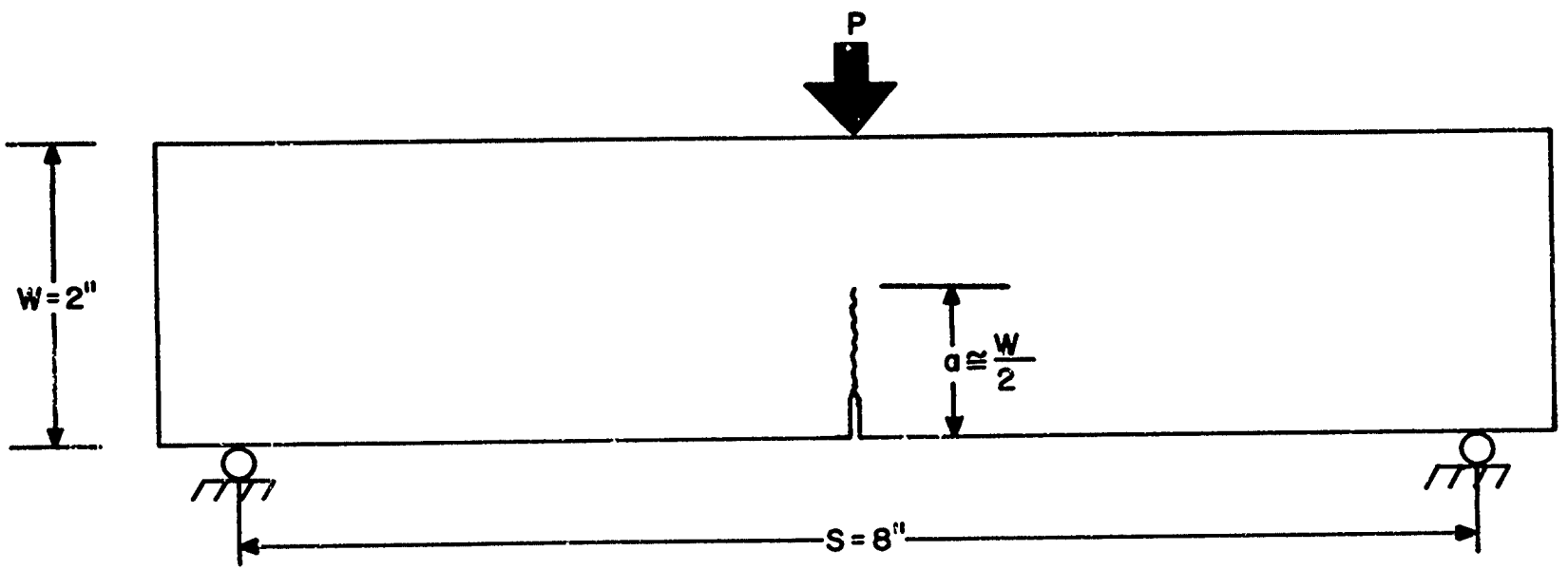
Table 3 - R-14 (0.06% oxygen) Experimental Results

I.D. No.	Thermal Treatment*	Hardness (R_c)	Est. Y.S. (ksi)	Hydrogen (ppm)	First Load (no cracking) K_I (ksi$\sqrt{in.}$)	First Load (no cracking) Time (hr)	Final Load (cracking) K_I (ksi$\sqrt{in.}$)	Final Load (cracking) Time (hr)	Comments
271	1700 Air 7FC	29.5	115	38	-	-	102	0	K_{Ix}
284	1700 Air 7FC	-	115	38†	57	26.8	85	5.6	cantilever
285	1700 Air 7FC	-	115	38†	66.5	70	76	17.5	cantilever
277	1700V7FC	28.0	110	9	-	-	121	0	K_{Ix}
278	1700V7FC	-	110	9†	0	0	81	9.1	3-point
313	1700V7FC+1700$H_2$7FC	27.5	110	33†	-	-	114	0	K_{Ix}
311	1700V7FC+1700$H_2$7FC	28.0	110	59†	-	-	87	0	K_{Ix}
312	1700V7FC+1700$H_2$7FC	-	110	59†	50	15.9	80	9.1	cantilever
283	1700V7FC+1700$H_2$7FC	27.5	110	124†	-	-	92	0	K_{Ix}
287	1700V7FC+1700$H_2$7FC	-	110	124†	75	5.3	85	13.0	3-point
298	1700V7FC+1700$H_2$7FC	-	110	124†	-	-	82	10.4	3-point

All K_{Ix} - K_{Ic} measurements 3-point bend. Maximum load reached in 10-15 seconds, at constant crosshead velocity of 0.005 in./sec.

* 1700V7FC = Annealed at 1700° F in vacuum for 7 hours, then furnace cooled, for example.

† Estimated hydrogen content. No analysis available.



B = THICKNESS = $\frac{W}{2}$

Fig. 1 – Sketch of Bend specimen

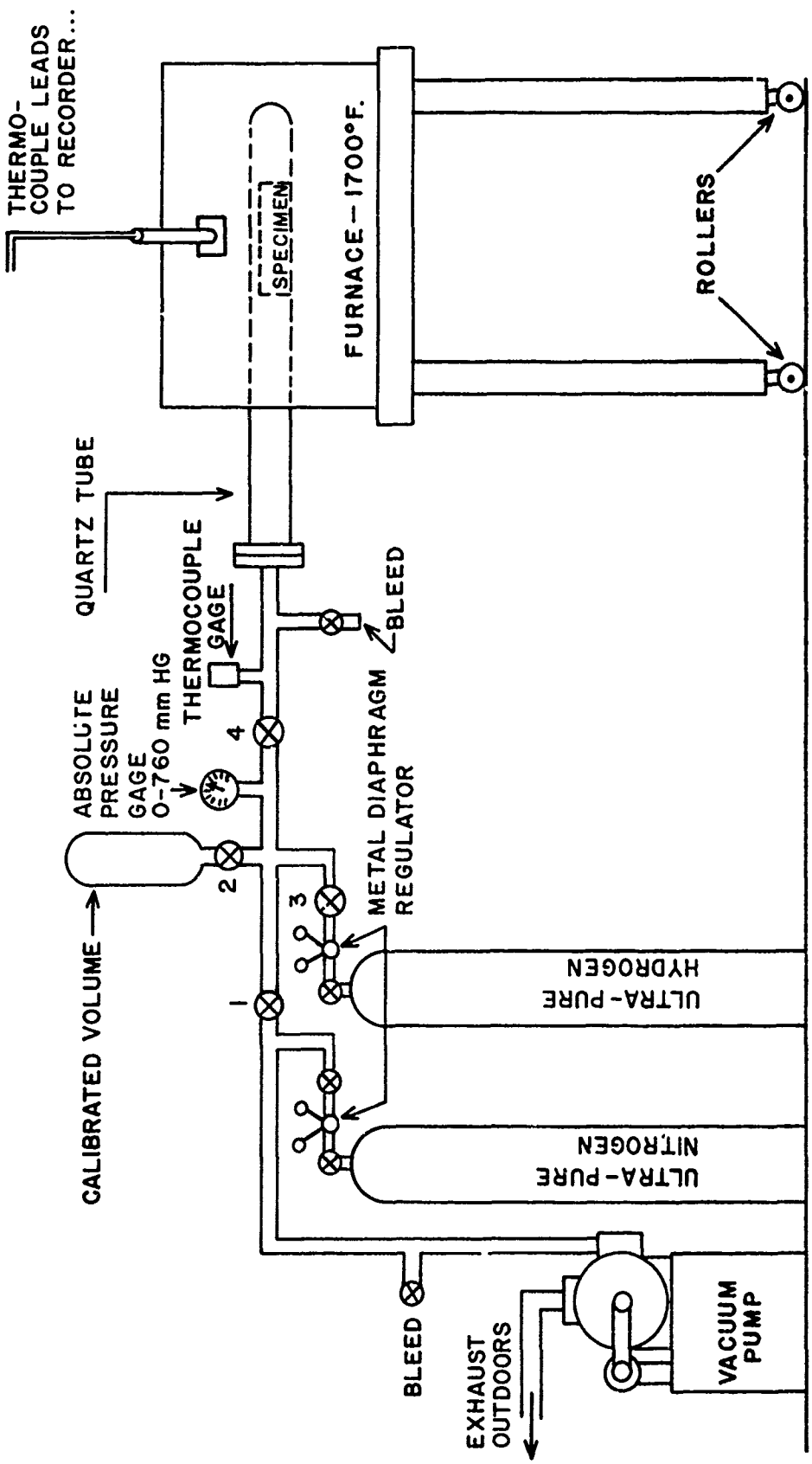
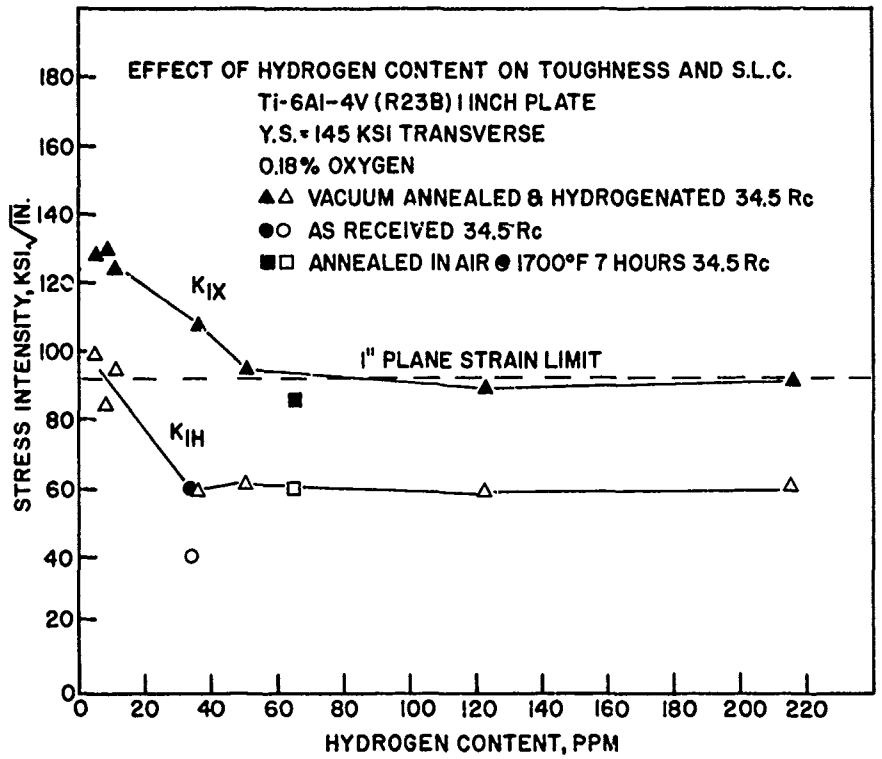
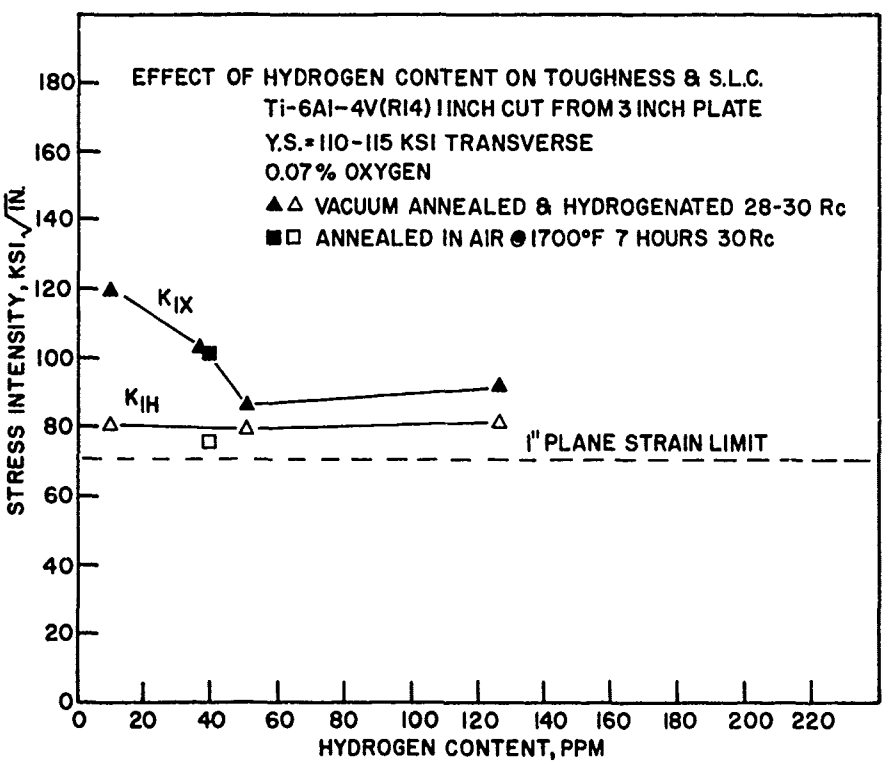


Fig. 2 – Sketch of hydrogenation apparatus



(a) - R-23B



(b) - R-14

Fig. 3 - Effect of Hydrogen content on K_{Ix} and K_{IH}.